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of

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for

***In-Situ* Oxidized Films for Use as Gap Layers for Spin Valve Sensor and  
Methods of Manufacture**

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1                                    ***In-Situ* Oxidized Films for Use as Gap Layers**  
2                                    **for a Spin-Valve Sensor and Methods of Manufacture**

3                                    **BACKGROUND OF THE INVENTION**  
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5        1.        **Related Applications**

6                    This application is a continuation-in-part of our co-pending patent application Serial  
7        No. 09/919280, filed on July 31, 2001 for *In-Situ* Oxidized Films for Use as Cap and Gap  
8        Layers in a Spin-Valve Sensor and Methods of Manufacture.  
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10       2.        **The Field of the Invention**

11                   The present invention relates to spin-valve sensors for reading information signals  
12        from a magnetic medium and more particularly to novel structures for spin-valve sensors and  
13        magnetic recording systems which incorporate such sensors.  
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15       3.        **The Relevant Art**

16                   Computer systems generally utilize auxiliary memory storage devices having  
17        magnetic media on which data can be written and from which data can be read for later use.  
18        A direct access storage device, such as a disk drive incorporating rotating magnetic disks, is  
19        commonly used for storing data in magnetic form on the disk surfaces. Data are recorded on  
20        concentric, radially spaced tracks on the disk surfaces. Magnetic recording heads carrying  
21        read sensors are then used to read data from the tracks on the disk surfaces.

22                   In high capacity disk drives, a giant magnetoresistance (GMR) head carrying a spin-  
23        valve sensor is now extensively used to read data from the tracks on the disk surfaces. This  
24        spin-valve sensor typically comprises two ferromagnetic films separated by an electrically  
25        conducting nonmagnetic film. The resistance of this spin-valve sensor varies as a function of  
26        the spin-dependent transmission of conduction electrons between the two ferromagnetic  
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1 films and the accompanying spin-dependent scattering which takes place at interfaces of the  
2 ferromagnetic and nonmagnetic films.

3 In the spin-valve sensor, one of the ferromagnetic films, referred to as a pinned layer,  
4 typically has its magnetization pinned by exchange coupling with an antiferromagnetic film,  
5 referred to as a pinning layer.

6 The magnetization of the other ferromagnetic film, referred to as a "sensing" or "free"  
7 layer is not fixed, however, and is free to rotate in response to the field from the magnetic  
8 medium (the signal field). In the spin-valve sensor, the GMR effect varies as the cosine of  
9 the angle between the magnetization of the pinned layer and the magnetization of the sensing  
10 layer. Recorded data can be read from a magnetic medium because the external magnetic  
11 field from the magnetic medium causes a change in the direction of magnetization in the  
12 sensing layer, which in turn causes a change in the resistance of the spin-valve sensor and a  
13 corresponding change in the sensed voltage.

14 Figure 1 shows a typical prior art GMR head 100 comprising a pair of end regions  
15 103 and 105 separated by a central region 102. The central region 102 is formed by  
16 depositing a spin-valve sensor 128 onto a bottom gap layer 118, which is previously  
17 deposited on a bottom shield layer 120, which is, in turn, previously deposited on a substrate.

18 Two end regions 103 and 105 abut the edges of the central region 102. In the spin-valve  
19 sensor 128, a ferromagnetic sensing layer 106 is separated from a ferromagnetic pinned layer  
20 108 by an electrically conducting nonmagnetic spacer layer 110. The magnetization of the  
21 pinned layer 108 is fixed through exchange coupling with an antiferromagnetic pinning layer  
22 114. This spin-valve sensor includes seed layers 116, on which the pinning, pinned, spacer  
23 and sensing layers of the spin-valve sensor 128 grow with preferred crystalline textures  
24 during sputtering so that desired improved GMR properties are attained. The end regions 103  
25 and 105 are also formed by depositing longitudinal bias (LB) and conducting lead layers 126  
26 on the bottom gap layer 118 and at the spin-valve sensor 128. The end regions 103, 105 abut  
27 the central region 102. The central and end regions are sandwiched between electrically

insulating nonmagnetic films, one referred as a bottom gap layer 118 and the other referred as a top gap layer 124.

The disk drive industry has been engaged in an ongoing effort to increase the recording density of the hard disk drive, and correspondingly to increase the overall signal sensitivity to permit the GMR head of the hard disk drive to read smaller changes in magnetic fluxes. The major property relevant to the signal sensitivity of a spin-valve sensor in the GMR head is its GMR coefficient. A higher GMR coefficient leads to higher signal sensitivity and enables the storage of more bits of information on a disk surface of a given size. The GMR coefficient of the spin-valve sensor is expressed as  $\Delta R_G/R_{//}$ , where  $R_{//}$  is a resistance measured when the magnetizations of the free and pinned layers are parallel to each other, and  $\Delta R_G$  is the maximum giant magnetoresistance (GMR) measured when the magnetizations of the free and pinned layers are antiparallel to each other.

In certain spin-valve sensors, particularly those with Co-Fe/Ni-Fe films as sensing layers 106, a cap layer 112 is often formed over the sensing layers. The cap layer 112 serves several purposes, and plays a crucial role in attaining a high GMR coefficient. For instance, a Cu cap layers is thought to induce spin filtering, while a Cu-O cap layer is thought to induce specular scattering. Both spin filtering and specular scattering are believed to increase the GMR coefficient of a spin-valve sensor. In addition, a cap layer may be employed to prevent the underlying sensing layers from interface mixing occurring immediately during depositions and oxygen diffusion occurring during subsequent annealing, thereby maintaining suitably soft magnetic properties of the sensing layer and improving the thermal stability of the spin-valve sensor. The term "soft magnetic property" refers to the capability of a spin-valve sensor to sense very small magnetic fields.

Currently, a Ta cap layer is used in many conventional spin-valve sensors. However, the Ta cap layer does not exhibit desired specular scattering, and is considered inadequate in preventing the sensing layers from interface mixing and oxygen diffusion. Interface mixing originates from direct contact between the sensing layers and the Ta cap layer, and causes a

substantial loss in the magnetic moment of the sensing layers. For one currently used spin-valve sensor with  $0.32 \text{ memu/cm}^2$  sensing layers, this magnetic moment loss accounts for 25% of the magnetic moment of the sensing layers. Oxygen diffusion originates from low passivity of the Ta cap layer, which oxidizes continuously and entirely during annealing, such that oxygen eventually penetrates into the sensing layers, causing more losses in the magnetic moment of the sensing layers.

Another limiting factor of the disk drive recording density is the dimensions of the GMR head. The recording density of the disk drive is inversely proportional to the total thickness of the spin-valve sensor, the gap layers 118 and 124. In other words, in order to increase the disk drive recording density the thicknesses of the spin-valve sensor, the gap layers 118 and 124 must be decreased. Several challenges have arisen in the miniaturization of the gap layers 118 and 124.

The primary duties of the gap layers 118 and 124 are to prevent electrical shorting between the spin-valve sensor 128 and the shield layers 120 and 130, and thus to ensure the functionality of the spin-valve sensor 128. In order to prevent this electrical shorting, a spin-valve sensor must be sandwiched between gap layers 118 and 124 of substantial thicknesses.

The gap layers 118 and 124 have been a limiting factor in the miniaturization of the GMR head 100, because as the thicknesses of the gap layers 118 and 124 decreases, the possibility of electrical shorting increases, causing the GMR head to be non-functional.

Thus, it can be seen from the above discussion that there is a need existing in the art for an improved spin-valve sensor with an increased GMR coefficient, and for improved gap layers with decreased thicknesses.

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In one embodiment, a bottom shield layer preferably formed of a Ni-Fe film and a bottom gap layer preferably formed of an  $\text{Al}_2\text{O}_3$  film are deposited on a wafer. Multiple seed layers preferably formed of  $\text{Al}_2\text{O}_3$ , Ni-Cr-Fe and Ni-Fe films are deposited on the bottom gap layer. A pinning layer preferably formed of a Pt-Mn film is then deposited on the multiple seed layers. Pinned layers preferably formed of Co-Fe, Ru and Co-Fe films are then deposited on the pinning layer. A spacer layer preferably formed of an oxygen-doped, *in-situ* oxidized Cu-O film is then deposited on the pinned layer. Sensing layers preferably formed of Co-Fe and Ni-Fe films are then deposited on the spacer layer. A cap layer preferably formed of an *in-situ* oxidized Al film (Al-O) is then formed on the sensing layer with a deposition/*in-situ* oxidation process. *In-situ* oxidation is preferred for attaining a high GMR coefficient.

The deposition/*in-situ* oxidation process preferably comprises depositing a metallic film in a vacuum in a DC magnetron sputtering module, and then conducting the *in-situ* oxidation for a wide range of time in a wide range of oxygen pressures in an oxidation module. In one embodiment given by way of example, the *in-situ* oxidation is conducted for a period of about 8 minutes in an oxygen gas of about 0.5 Torr. In another embodiment given by way of example, the *in-situ* oxidation is conducted for a period of about 4 minutes in an oxygen gas of about 2 Torr. The exposure to oxygen is preferably conducted with a moderate temperature, such as ambient room temperature.

The top and bottom gap layers are preferably deposited using the deposition/*in-situ* oxidation process of the present invention. In order to achieve designed thicknesses, multiple layers may be alternatively deposited and *in-situ* oxidized using the deposition/*in-situ* oxidation process. Preferably, when forming the top and bottom gap layers, the alternating oxidized layers are fully oxidized.

These and other objects, features, and advantages of the present invention will become more fully apparent from the following description and appended claims, or may be learned by the practice of the invention as set forth hereinafter.

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### BRIEF DESCRIPTION OF THE DRAWINGS

In order that the manner in which the advantages and objects of the invention are obtained will be readily understood, a more particular description of the invention briefly described above will be rendered by reference to specific embodiments thereof which are illustrated in the appended drawings. Understanding that these drawings depict only typical embodiments of the invention and are not therefore to be considered to be limiting of its scope, the invention will be described and explained with additional specificity and detail through the use of the accompanying drawings in which:

Figure 1 is a cross-sectional view illustrating the structure of a GMR head of the prior art;

Figure 2 is a schematic block diagram illustrating a magnetic recording disk drive system;

Figure 3 is a cross-sectional view illustrating the structure of a GMR head in accordance with one embodiment of the present invention;

Figure 4 is a schematic block diagram illustrating the structure of a cap layer in one embodiment of the present invention;

Figure 5 is a schematic block diagram illustrating a plurality of *in-situ* oxidized Al films in an alternative embodiment of the present invention;

Figure 6 is a schematic block diagram illustrating an integrated DC magnetron/ion beam sputtering system suitable for use with the present invention;

Figure 7 is a schematic flow chart illustrating a method for deposition/*in-situ* oxidization of the present invention;

1           Figure 8 is a cross-sectional view illustrating the structure of TMR head in  
2 accordance with one embodiment of the present invention; and

3           Figure 9 is a chart showing the sheet resistance ( $R_s$ ) versus the *in-situ* oxidization  
4 time ( $t$ ) for  $\text{Al}_2\text{O}_3/\text{Ni-Fe}$  and  $\text{Al}_2\text{O}_3/\text{Ni-Fe/Al}$  films.  
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control signals and internal clock signals. Typically, the control unit 229 comprises logic control circuits, storage means, and a microprocessor. The control unit 229 generates control signals to control various system operations such as drive motor control signals on a line 223 and head position and seek control signals on a line 228. The control signals on the line 228 provide the desired current profiles to optimally move and position the slider 213 to the desired data track on the magnetic disk 212. Read and write signals are communicated to and from the read/write heads 221 by means of a recording channel 225. In the depicted embodiment, the read/write heads 221 incorporate a spin-valve sensor of the present invention.

Referring now to Figure 3, shown therein is one embodiment of a GMR head 300 employing the *in-situ* oxidized films of the present invention. While the GMR head 300 is discussed in relation to one example of a GMR sensor, the present invention is also applicable to other GMR sensors, the construction of which should be readily apparent from the present discussion.

The depicted GMR head 300 includes a bottom-type synthetic spin-valve sensor, but of course, other types of spin-valve sensors may also be formed using the novel *in-situ* process of the present invention. The depicted GMR head 300 comprises a pair of end regions 303, 305 separated by a central region 301. The central region 301 is formed by depositing the spin-valve sensor 328 on a bottom gap layer 304. The end regions 303 and 305 are formed by depositing longitudinal bias (LB) and conducting lead layers 326 on a bottom gap layer 304. The end regions 303, 305 abut the central region 301.

In accordance with one embodiment, ferromagnetic sensing layers 307 (often collectively referred to as "free layers") are shown separated from ferromagnetic pinned layers 309 by a spacer layer 316. In one embodiment, the spacer layer 316 is a non-magnetic 22 Å thick, electrically-conducting, oxygen-doped, *in-situ* oxidized Cu (Cu-O) film. Under one embodiment of the present invention, the sensing layers 307 comprise a 9 Å thick Co-Fe film 318 and an adjacent 27 Å thick Ni-Fe film 320. The pinned layer 309 comprises a 20 Å

1 thick Co-Fe film 310, an 8 Å thick Ru film 312, and a 22 Å thick Co-Fe film 314. The  
2 magnetizations of the pinned layers 309 are fixed through exchange coupling with a 200 Å  
3 thick antiferromagnetic Pt-Mn pinning layer 308.

4 One manner of forming the GMR head 300 of Figure 3 will be discussed herein by  
5 way of example. In the depicted embodiment, a bottom shield layer 311, preferably formed  
6 of a 1 µm thick Ni-Fe film, and a bottom gap layer 304, preferably formed of a 100 Å thick  
7 Al<sub>2</sub>O<sub>3</sub> film, are deposited on a wafer 302. Seed layers 306 may then be deposited on the  
8 bottom gap layer 304.

9 In one embodiment, multiple seed layers 306, preferably formed of a 30 Å thick  
10 Al<sub>2</sub>O<sub>3</sub> film, a 30 Å thick Ni-Cr-Fe film, and a 10 Å thick Ni-Fe film, are deposited on the  
11 bottom gap layer 304. The 100 Å thick Al<sub>2</sub>O<sub>3</sub> film used as the bottom gap layer 304 may be  
12 sputter-deposited in an argon gas from an Al<sub>2</sub>O<sub>3</sub> target, while the 30 Å thick Al<sub>2</sub>O<sub>3</sub> film used  
13 as the seed layer is preferably reactively sputter-deposited from an Al target in mixed gases  
14 of argon and oxygen. The multiple seed layers are used to provide a desirable surface with a  
15 strong face-centered-cubic (FCC) {111} crystalline texture and coarse polycrystalline grains  
16 on which the remaining layers of the spin-valve sensor may be grown epitaxially in order to  
17 attain a high GMR coefficient.

18 An antiferromagnetic pinning layer 308, preferably formed of a 200 Å thick Pt-Mn  
19 film, is then, under this embodiment, deposited on the seed layers 306. The pinned layers  
20 309, preferably formed of a 20 Å thick Co-Fe film, an 8 Å thick Ru film, and a 22 Å thick  
21 Co-Fe film, are then deposited on the pinning layer 308. A spacer layer 316, preferably  
22 formed of an oxygen-doped, *in-situ* oxidized 22 Å thick Cu (Cu-O) film, is then deposited on  
23 the pinned layer 309. Sensing layers 307 are then formed, preferably of a 9 Å thick Co-Fe  
24 film 318 and a 27 Å thick Ni-Fe film 320, on the spacer layer 316. A cap layer 322,  
25 preferably formed of an *in-situ* oxidized Al film (Al-O), is then formed on the sensing layers  
26 307. The cap layer 322 is preferably formed with the *in-situ* deposition/oxidization process  
27 of the present invention, one example of which is discussed below with respect to Figure 7.

The cap layer 322 may be formed of any suitable *in-situ* oxidized film, examples of which include *in-situ* oxidized Al, Hf, Si, Y, and Zr films. An *in-situ* oxidized Al film is considered to be particularly effective, due to its amorphous state which is thought to promote high specular scattering.

In addition, a metallic film may be disposed between the sensing layer 307 and the cap layer 322. In one embodiment, the metallic film comprises a Cu film with a thickness in a range of between about 6 Å and about 12 Å.

A top gap layer 324 is then preferably formed on the cap layer 322. In one embodiment, the top gap layer 324 is a 100 Å thick Al<sub>2</sub>O<sub>3</sub> film. A top shield layer 325 may then be formed on top of the top gap layer 324. In one embodiment, the top shield layer 325 is formed of a 1 µm thick Ni-Fe film.

The low-passivity Ta film of the prior art cap layer (112 of Figure 1) oxidizes continuously and entirely. In contrast, using the deposition/*in-situ* oxidation process of the present invention, a high-passivity film is oxidized only at its surface, resulting in a natural oxide cap layer which is dense and highly protective against oxygen diffusion into the sensing layer.

The cap layer of the present invention is preferably only partially oxidized, with the upper portion of the film oxidized, while the lower portion of the film directly adjacent the sensing layer remains intact and substantially free from oxygen. The natural dense oxide layer formed in the upper portion of the film is thought to provide higher specular scattering than a thick oxide film either sputtered from an oxide target or reactively sputtered from a metallic target in mixed gases of argon and oxygen. This high specular scattering causes a substantial increase in the GMR coefficient of the spin-valve sensor.

In addition, severe interface mixing between the sensing layer and a conventional Ta cap layer is substantially reduced, and oxygen penetration from the conventional Ta layer into the sensing layer is prevented. As a result, the soft magnetic properties of the sensing layers are substantially improved. In the preferred embodiments, the *in-situ* oxidized film is

Figure 5 illustrates an alternative embodiment in which repeated deposition/*in-situ* oxidation processes are used to form the top and bottom gap layers 324 and 304. The top and bottom gap layers 324 and 304 may be formed of a plurality of *in-situ* oxidized metallic Al films 400. The deposition/*in-situ* oxidation process is repeated until

1 designed thicknesses of the top and bottom gap layers 324 and 304 are attained. Each *in-situ*  
2 oxidized film 400 is formed in the manner described with respect to Figure 4. While reactive  
3 pulsed-DC magnetron sputter-deposition of an  $\text{Al}_2\text{O}_3$  film is preferred for the first seed layer  
4 (which can also be treated as part of the bottom gap layer 304), full *in-situ* oxidation is  
5 preferred for the bottom gap layer 304 to ensure high breakdown voltages.

6 While *in-situ* partial oxidation is preferred for the cap layer (which can also be treated  
7 as part of the top gap layer 324), full *in-situ* oxidation is also preferred for the top gap layer  
8 324 to also ensure high breakdown voltages. Hence in one example, to form each *in-situ*  
9 oxidized film as part of the top and bottom gap layer 324 and 304, an Al film is deposited  
10 and *in-situ* oxidized for 16 min in an oxygen gas of 2 Torr. In another example, the Al film  
11 is oxidized for a longer time in the oxygen gas with a higher pressure. The deposition/*in-situ*  
12 oxidation process is repeated until the designed thickness is attained. In one example, each  
13 *in-situ* oxidized film has a thickness of about 10 Å and 10 layers are deposited. In a further  
14 example, up to 20 layers are deposited.

15 Referring now to Figure 6, shown therein is one embodiment of an integrated DC  
16 magnetron/ion beam sputtering system 600 suitable for fabricating the GMR head 300 and  
17 for conducting the deposition/*in-situ* oxidation process of the present invention. The  
18 sputtering system 600 of Figure 6 is sold by the Veeco Corporation of Plainview, New York.

19 The sputtering system 600 as depicted comprises a transport module 602 surrounded by a  
20 first single-target DC magnetron sputtering module 604, a multi-target DC magnetron  
21 sputtering module 606, a multi-target ion beam sputtering module 608, and a second single-  
22 target DC magnetron sputtering module 610. Two loadlocks 616 allow the ingress and  
23 egress of wafers. A control panel 614 controls the parameters and processes of the sputtering  
24 system 600.

25 Figure 7 is a schematic flow chart illustrating one embodiment of a method 700 of  
26 forming *in-situ* oxidized films of the present invention. The method 700 starts 702 and a  
27 metallic film is formed by sputter-deposition 704, preferably using an integrated DC-

1 magnetron/ion beam sputtering system, such as that described with reference to Figure 6.  
2 Under the preferred embodiment of the present invention, the sputter-deposition 704 of the  
3 metallic film is accomplished in an atmosphere of argon gas of 3 mTorr.

4 Once the desired thickness of the metallic film on the wafer has been achieved, the  
5 wafer is moved in a vacuum 706 through the transport module 602 to the single-target DC  
6 magnetron module 604 or 610, which can be used as an *in-situ* oxidization module. The  
7 metallic film 504 is then *in-situ* oxidized in the oxidization module 604 or 610, where the  
8 oxygen gas is introduced 708. In one embodiment, the pressure of the oxygen gas in the  
9 oxidization module 604 or 610 is in a range of about 0.5 to about 10 Torr. To ensure full *in-*  
10 *situ* oxidation, the pressure of the oxygen gas in the *in-situ* oxidation module 604 or 610 is  
11 preferably 2 Torr or greater. The temperature is preferably maintained at about room  
12 temperature (i.e., about 70 ° F).

13 The full *in-situ* oxidation of the embodiment of Figure 5 is conducted for a period of  
14 time of about 16 min in an oxygen gas of 2 Torr. The *in-situ* oxidization is preferably a  
15 natural oxidization performed at ambient room temperature. When used to form multilayer  
16 films such as the embodiment of Figure 5, the method 700 is repeated, until determining 710  
17 that the selected number of layers or designed thickness has been reached. The method 700  
18 then ends 712.

19 In one embodiment, the sputtering system 600 and deposition/*in-situ* oxidation  
20 method 700 are used in the fabrication of a GMR head with cap and gap layers all formed of  
21 *in-situ* oxidized Al films. In this embodiment, the read gap thickness is designed to be as  
22 small as 600 Å for magnetic recording at ultrahigh densities ( $\geq 30$  Gb/in<sup>2</sup>). To attain this read  
23 gap thickness, the spin-valve sensor of the GMR head is sandwiched between 100 Å thick  
24 Al-O top and bottom gap layers.

25 The bottom gap layer 304 formed of 10 layers of *in-situ* oxidized Al films are formed  
26 on a wafer in the first single-target DC magnetron sputtering module 604. The deposition/*in-*  
27 *situ* oxidation process is repeated 10 times until a 100 Å thick bottom gap layer is attained.

To ensure full *in-situ* oxidization that is preferred for attaining a high breakdown voltage, the deposition of an Al film with DC magnetron sputtering from a pure Al target in an argon gas of 3 mTorr and the subsequent *in-situ* oxidation for 16 minutes in an oxygen gas of 2 Torr are conducted alternatively for a total of 10 times.

After the formation of the bottom gap layer 304 in the first single-target DC magnetron sputtering module 604, the wafer is then transferred to the second single-target DC magnetron sputtering module 610 for the deposition of the first seed layer that is in one embodiment formed of a 30 Å thick Al<sub>2</sub>O<sub>3</sub> film. The Al<sub>2</sub>O<sub>3</sub> film is deposited, preferably with reactive pulsed-DC magnetron sputtering from a pure Al target in mixed argon and oxygen gases of 2.25 and 0.75 mTorr, respectively. This Al<sub>2</sub>O<sub>3</sub> film when used as a seed layer in disclosed embodiments plays a significant role in increasing the {111} crystalline textures of subsequently deposited films and in improving the GMR coefficient of the spin-valve sensor. In contrast, an Al-O film, formed with the deposition/*in situ* oxidation process, has been found to be a non-suitable seed layer.

The wafer is then transferred to the multi-target ion beam sputtering module 608 for the deposition of the second and third seed layers that may be formed of a 30 Å thick Ni-Fe-Cr film and a 10 Å thick Ni-Fe film, respectively. The Ni-Cr-Fe and Ni-Fe films are preferably deposited in a xenon gas of 0.12 mTorr.

The wafer is then transferred to the multi-target DC magnetron sputtering module 606 for the deposition of the remaining layers of the spin-valve sensor, in one embodiment including a 200 Å thick Pt-Mn film, a 20 Å thick Co-Fe film, an 8 Å thick Ru film, a 22 Å thick Co-Fe film, an oxygen-doped/*in-situ* oxidized 22 Å thick Cu (Cu-O) film, a 9 Å thick Co-Fe film, a 27 Å thick Ni-Fe film, and a 10 Å thick *in-situ* oxidized Al (Al-O) film. All the metallic films except the Cu-O film are deposited in an argon gas of 3 mTorr and in a magnetic field of 40 Oe parallel to an alignment mark. To form the Cu-O film, a Cu film is deposited in mixed argon and oxygen gases of 2.985 and 0.015 mTorr, respectively, and then *in-situ* oxidized in mixed argon and oxygen gases of 2.94 and 0.06 mTorr for 4 minutes,

1 respectively. To form the Al-O film, an Al film is deposited in an argon gas of 3 mTorr and  
2 then *in-situ* oxidized in an oxygen gas of 0.5 Torr for 8 minutes.

3 In addition, to form an additional Cu-O or Ru-O film before the formation of the Al-  
4 O film, a Cu or Ru film is deposited in an argon gas of 3 mTorr and then *in-situ* oxidized in  
5 an oxygen gas of 0.5 Torr for 8 minutes.

6 After the depositions, the wafer is annealed for 300 minutes at 265 °C in a magnetic  
7 field of 10 kOe perpendicular to an alignment mark. After annealing, a 30 Å thick Ta film is  
8 deposited for the use as an adhesion layer for photoresist layers applied in the subsequent  
9 patterning process. In this patterning process, bilayer photoresists are applied and exposed in  
10 a photolithographic tool to mask the spin-valve sensor in a central region, and then  
11 developed in a solvent to form an undercut. The spin-valve sensor in unmasked end regions  
12 is removed by ion milling until the Al<sub>2</sub>O<sub>3</sub> first seed layer is exposed, and longitudinal bias  
13 (LB) and first conducting leads (LD<sub>1</sub>) layers are immediately deposited.

14 Subsequently, the bilayer photoresists are lifted off and a similar patterning process  
15 continues for the deposition of second conducting leads (LD<sub>2</sub>) layers. Ion milling or reactive  
16 ion etching is applied to remove the 30 Å thick Ta film. The top gap layer-formed of 10  
17 layers of *in-situ* oxidized Al films is then formed on the wafer in the first single-target DC  
18 magnetron sputtering module 604. The deposition/*in-situ* oxidation process is repeated 10  
19 times until about a 100 Å thick top gap layer is attained. To ensure full *in-situ* oxidization  
20 preferred for attaining high breakdown voltage, the deposition of O) film, a 9 Å thick Co-Fe  
21 film, a 27 Å thick Ni-Fe film, and a 10 Å thick *in-situ* oxidized Al (Al-O) film. All the  
22 metallic films except the Cu-O film are deposited in an argon gas of 3 mTorr and in a  
23 magnetic field of 40 Oe parallel to an alignment mark. To form the Cu-O film, a Cu film is  
24 deposited in mixed argon and oxygen gases of 2.985 and 0.015 mTorr, respectively, and then  
25 *in-situ* oxidized in mixed argon and oxygen gases of 2.94 and 0.06 mTorr for 4 minutes,  
26 respectively. To form the Al-O film, an Al film is deposited in an argon gas of 3 mTorr and  
27 then *in-situ* oxidized in an oxygen gas of 0.5 Torr for 8 minutes.

Table 1 lists magnetic and magnetoresistive properties of spin-valve sensors used in the prior art and in this invention.

Seed Layer	Ta	Al <sub>2</sub> O <sub>3</sub> /Ni-Cr-Fe/Ni-Fe	Al <sub>2</sub> O <sub>3</sub> /Ni-Cr-Fe/Ni-Fe	Al <sub>2</sub> O <sub>3</sub> /Ni-Cr-Fe/Ni-Fe
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Cap Layer	Ta	Al-O	Cu-O/Al-O	Ru-O/Al-O
$m_1$ (memu/cm <sup>2</sup> )	0.28	0.32	0.32	0.32
$\lambda_s$ ( $\times 10^{-6}$ )	-0.3	+0.15	-1.2	-1.58
$H_C$ (Oe)	14.3	6.2	6.2	6.2
$H_F$ (Oe)	-25.3	-6.6	-12.1	-12.5
$R_{  }$ ( $\Omega$ )	20.2	16.6	15.6	16.4
$\Delta R_G/R_{  }$ (%)	7.7	13.8	13.4	13.4
$\Delta R_G$ ( $\Omega$ )	1.56	2.29	2.09	2.20

The replacement of the Ta cap layer with the *in-situ* oxidized cap layer causes an increase in the areal magnetic moment of the sensing layer ( $m_1$ ) from 0.28 to 0.32 memu/cm<sup>2</sup>, a decrease in the amplitude of the ferromagnetic coupling field ( $|H_F|$ ) from 25.3 to 6.6 Oe, and a decrease in the easy-axis coercivity ( $H_C$ ) from 14.3 to 6.2 Oe. These changes in magnetic properties may originate from substantially reduced interface mixing at the interface between the sensing and *in-situ* oxidized cap layers. The replacements of the Ta seed and Ta cap layers with the Al<sub>2</sub>O<sub>3</sub>/Ni-Cr-Fe/Ni-Fe seed and Al-O cap layers, respectively, cause a decrease in the sheet resistance of the spin-valve sensor ( $R_{||}$ ) from 20.2 to 16.6  $\Omega$ , but an increase in the GMR coefficient ( $\Delta R_G/R_{||}$ ) from 7.7 to 13.8 %. These changes in magnetoresistive properties may originate from grain coarsening in the spin-valve sensor due to recrystallization in the Ni-Cr-Fe/Ni-Fe films, and improved specular scattering at the interface between the sensing and *in-situ* oxidized cap layers.

In addition, the sandwiching of the Cu-O or Ru-O cap layer between the sensing and Al-O cap layers causes a slight decrease in  $\Delta R_G/R_{||}$ , and a transition in the saturation magnetostriction of the sensing layer ( $\lambda_s$ ) from positive to negative values. This transition may originate from an enforced metallic contact between the sensing and cap layers. This negative  $\lambda_s$  is preferred and is typically controlled in a range from  $-2 \times 10^{-6}$  to

1  $-1 \times 10^{-6}$  for improving magnetic and thermal stability of the spin-valve sensor. Hence, in  
2 spite of the fact that the use of the Cu-O or Ru-O cap layer causes the slight decrease in  
3  $\Delta R_G/R_{||}$ , it may nevertheless be employed for improving magnetic and thermal stability of  
4 the spin-valve sensor.

5 The *in-situ* oxidation process of the present invention may also be applied to other  
6 types of read heads, one example of which includes, a tunneling magnetoresistance  
7 (TMR) head comprising a magnetic-tunnel-junction sensor. The TMR sensor is well  
8 known in the art and shares a similar structure to the GMR head, discussed by way of  
9 example herein. Nevertheless, the application of the *in-situ* formed gap layers of the  
10 present invention is also applicable to the TMR sensor, as will be discussed. While one  
11 example of a TMR head having a TMR sensor of the prior art will be discussed here, the  
12 present invention likewise applies to other TMR heads, the construction of which should  
13 be readily apparent from the present discussion.

14 Referring to Figure 8, shown therein is a TMR head 800 comprising a TMR  
15 sensor 828 and a longitudinal bias (LB) stack 830 in a central region 801. Figure 8 shows  
16 the lower portion of the TMR sensor and an insulating gap layer 826 in each of a pair of  
17 end regions 803. The TMR sensor 828, the longitudinal bias (LB) stack 830, and the gap  
18 layer 826 are formed by a suitable method such as DC-magnetron or ion-beam sputter-  
19 deposition on a wafer 802.

20 In the TMR sensor 828, ferromagnetic sensing layers 807 are separated from  
21 ferromagnetic transverse pinned layers 809 by a 6 Å thick, in-situ oxidized, Al (Al-O)  
22 barrier layer 812. Under one embodiment of the present invention, the sensing layers 807  
23 comprise a 9 Å thick Co-Fe film 814 and an adjacent 27 Å thick Ni-Fe film 816. The  
24 transverse pinned layers 809 comprise an 18 Å thick Co-Fe film 806, an 8 Å thick Ru  
25 film 808 and a 24 Å thick Co-Fe layer film 810. The magnetizations of the transverse  
26 pinned layer 809 are fixed through exchange coupling with a 200 Å thick  
27

antiferromagnetic Pt-Mn transverse pinning layer 804 in a transverse directional perpendicular to an air bearing surface (ABS).

In the LB stack 830, a ferromagnetic 30 Å thick Co-Fe longitudinal pinned layer 820 is separated from the ferromagnetic Co-Fe/Ni-Fe sense layers 807 by a 30 Å thick nonmagnetic Ru decoupling layer 818, and is overlaid with a 60 Å thick antiferromagnetic Ir-Mn longitudinal pinning layer 822 and a nonmagnetic 90 Å thick Ru cap layer 824. The magnetization of the longitudinal pinned layer 820 are fixed through exchange coupling with the antiferromagnetic Ir-Mn longitudinal pinning layer 804 in a longitudinal directional parallel to the ABS.

In one embodiment of a manner of forming the TMR head 800, a bottom shield layer 802, preferably formed of a 1 µm thick Ni-Fe film, and a seed layer 811, preferably formed of a 90 Å thick Ta film are deposited on a wafer 801. An antiferromagnetic pinning layer 804, preferably formed of a 200 Å thick Pt-Mn film is deposited on the seed layers 811.

The pinned layers 809 are deposited on the pinning layer 804 and a barrier layer 812, preferably formed of an in-situ oxidized 6 Å thick Al (Al-O) film is deposited on the pinned layers 809. The sensing layers 807 are deposited on the barrier layer 812.

The LB stack, preferably formed of a 30 Å thick Ru film, a 40 Å thick Co-Fe film, a 60 Å thick Ir-Mn film, a 90 Å thick Ru film, and a 30 Å thick Ta film are then deposited on the sensing layer 807. All the depositions are preferably conducted with DC magnetron sputtering.

The LB stack in the central region plays a crucial role in achieving sensor stability, attaining high signal sensitivity, attaining high read efficiency, and eliminating side reading. The sensor stability can be easily achieved due to magnetostatic interactions between magnetic moments of the sense and the longitudinal pinned layers, which form a flux closure after the TMR sensor 828 overlaid with the LB stack 830 is ion-milled for the definition of the sensor width. The high signal sensitivity can be attained since the

1 magnetic moment of the longitudinal pinned layer is only needed to be 1.5 times of the  
2 magnetic moments of the sense layers for sensor stability, instead of more than 6 times  
3 when Cr/Co-Pt-Cr films are used in the first embodiment. The high read efficiency can be  
4 attained since stray fields stemming from the Co-Pt-Cr film used in the first embodiment  
5 do not exist at sensor edges. As a result, stray-field induced sensor stiffness at sensor  
6 edges is substantially reduced. Side reading can be eliminated since the TMR sensor and  
7 the LB stack are self-aligned in the fabrication process. As a result, a precise read width  
8 control can be achieved.

9 After the depositions, the wafer is annealed for 300 minutes at 265 °C in a  
10 magnetic field of 10 kOe perpendicular to an alignment mark, and then annealed again for  
11 20 minutes at 240 °C in a magnetic field of 200 Oe parallel to the alignment mark. These  
12 two anneals After the depositions, the wafer is annealed for 300 min at 265 °C in a  
13 magnetic cause the Pt-Mn film to pin the magnetizations of the Co-Fe/Ru/Co-Fe films in  
14 a direction perpendicular to the alignment mark, and cause the Ir-Mn film to pin the  
15 magnetization of its underlying Co-Fe film in a direction parallel to the alignment mark.  
16 After these two anneals, bilayer photoresists are applied and exposed in a  
17 photolithographic tool to mask the magnetic-tunnel-junction sensor TMR sensor 828 and  
18 the LB stack 830 in a read the central region, and then developed in a solvent to form an  
19 undercut.

20 Unmasked TMR sensor 828 and the LB stack 830 in unmasked end regions 803,  
21 805 of the GMR head 800 are removed by ion milling until the Al-O barrier layer is  
22 exposed. Subsequently, a gap region 826 is formed by depositing a plurality of layers,  
23 preferably formed of in-situ oxidized Al (Al-O), on the exposed Al-O film. In one  
24 embodiment, 22 layers are formed. The repeated deposition/in-situ oxidization process is  
25 preferably identical to that described above for Figure 7. After this repeated  
26 deposition/in-situ oxidization process, the bilayer photoresist mask is lifted off, and a  
27 further patterning process is applied to the wafer for opening the central region. Ion

milling or reactive ion etching is applied to remove the 30 Å thick Ta film and a top shield layer 825, preferably formed of a 1 µm thick Ni-Fe film, is deposited in the central region.

It should be noted that a partial in-situ oxidation is preferred for the formation of the cap layer of the spin-valve sensor, an optimal in-situ oxidation is preferred for the formation of the barrier layer of the magnetic-tunnel-junction sensor, and a full in-situ oxidation is preferred for the formation of the gap layers of the GMR head 700 and the TMR head 800. The partial in-situ oxidation may be attained after deposition of a 8 Å thick Al film and in-situ oxidation for 8 minutes in an oxygen gas of 0.5 Torr. The optimal in-situ oxidation may be attained after deposition of a 5.6 Å thick Al film and in-situ oxidation for 4 minutes in an oxygen gas of 2 Torr. Likewise, the full in-situ oxidation may be attained after deposition of a 8 Å thick Al film and in-situ oxidation for 16 minutes in an oxygen gas of 2 Torr.

In conducting the in-situ oxidization operations of the above embodiments, the Al films are preferred to be as thin as possible. A thicker Al film may be used and may be exposed to air for full ex-situ oxidization, but this full ex-situ oxidization is less preferred, as it may cause air contamination.

To determine suitable oxygen pressures and in-situ oxidization time needed for the full in-situ oxidization, the in-situ oxidation process is preferably monitored by in-situ probing. In so doing, Al<sub>2</sub>O<sub>3</sub>(30)/Ni-Fe(20) and Al<sub>2</sub>O<sub>3</sub>(30)/Ni-Fe(20)/Al(10) films are deposited on a glass substrate in DC magnetron sputtering modules, and are in-situ probed either during staying in a vacuum of  $4 \times 10^{-8}$  Torr or during in-situ oxidization in an oxygen gas of 2 Torr in an oxidation module, and then capped with a 60 Å thick Ta film.

Fig. 9 shows the sheet resistance ( $R_s$ ) attained under the present invention versus the in-situ oxidization time (t). Initial probing reveals that the Al<sub>2</sub>O<sub>3</sub>/Ni-Fe/Al films exhibit an

1  $R_s$  higher than the  $Al_2O_3/Ni-Fe$  films. This probing result is somewhat surprising, since both  
2 the low electrical resistivity of the Al film ( $6.0 \mu\Omega\text{-cm}$ ), which is much lower than of the Ni-  
3 Fe film ( $23.5 \mu\Omega\text{-cm}$ ), and an increase in the total film thickness should theoretically only  
4 cause a substantial decrease in  $R_s$ . However, it is believed by the inventors that atomic  
5 mixing at the interface between the Ni-Fe and Al films may occur, and thus cause a more  
6 substantial increase in  $R_s$ . Subsequent probing in the vacuum of  $4 \times 10^{-8}$  Torr reveals  
7 continuing increases in  $R_s$  for both the  $Al_2O_3/Ni-Fe$  and  $Al_2O_3/Ni-Fe/Al$  films. In-situ  
8 oxidization may thus gradually occur even in a vacuum where some oxygen gases still exist.

9 The  $Al_2O_3/Ni-Fe/Al$  films are also preferably in-situ probed in the vacuum of  $4 \times 10^{-8}$   
10 Torr for the first 30 seconds, and then in an oxygen gas of 2 Torr. The  $R_s$  substantially  
11 increases to a maximum value as soon as the oxygen gas is introduced, and then begins to  
12 decrease. This probing result is also somewhat surprising, since oxygen penetration during  
13 continuing in-situ oxidization should only cause an increase in  $R_s$ . It is believed by the  
14 inventors that this decrease in  $R_s$  indicates full in-situ oxidization, as described below. After  
15 introducing the oxygen gas, the Al film becomes immediately oxidized, and when the Al  
16 atoms at the interface between the Ni-Fe and Al films interface become oxidized, a sharp  
17 interface between the Ni-Fe and Al-O films may be created. As a result, specular electron  
18 scattering may occur in the Ni-Fe film confined by the  $Al_2O_3$  and Al-O films, leading to the  
19 decrease in  $R_s$ .

20 To determine suitable oxygen pressures and *in-situ* oxidization time needed for the  
21 full *in-situ* oxidization, in addition to the  $R_s$ , the magnetic moment of the underlying Ni-Fe  
22 film may also be monitored. The  $Al_2O_3/Ni-Fe/Al-O/Ta$  films are preferably annealed for 300  
23 minutes at  $265^\circ\text{C}$  in a direction parallel to the easy axis of the Ni-Fe film, and magnetic  
24 properties of the Ni-Fe film are measured with a high-sensitivity vibrating sample  
25 magnetometer. If the desired full *in-situ* oxidization is not attained yet, residual Al atoms in  
26 contact with the Ni-Fe film may diffuse into the Ni-Fe film during annealing, causing a loss  
27 in the magnetic moment of the Ni-Fe film from  $0.16 \text{ memu/cm}^2$  (equivalent to  $20 \text{ \AA}$  in Ni-Fe